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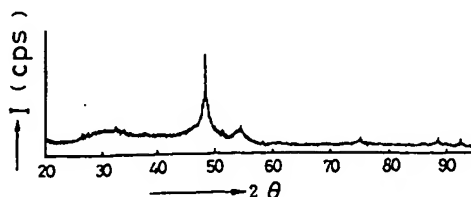
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table 1, 4th line : AL 8 by Fe 4 should read  
AL8DyFe4 has been filed pursuant to Rule 88  
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**20**  
**D-81635 München (DE)**(54) **High strength aluminum alloy.**

(57) An aluminum alloy includes an aluminum matrix, and an intermetallic compound  $Al_8REFe_4$  (wherein RE is at least one element selected from the group consisting of Y, La, Ce, Pr, Nd, Sm, Gd and Dy) having a grain size of at most  $1.0 \mu m$  and dispersed in the aluminum matrix. The volume fraction  $V_f$  of  $Al_8REFe_4$  is in a range of  $13\% \leq V_f \leq 38\%$ . In this way, if the volume fraction  $V_f$  of  $Al_8REFe_4$  is set at a value equal to or more than 13% ( $13\% \leq V_f$ ), the Vickers hardness  $H_v$  of the aluminum alloy can be increased to increase the strength. In addition, an increase in toughness of the aluminum alloy can be achieved by setting the volume fraction  $V_f$  of  $Al_8REFe_4$  at a value equal to or less than 38% ( $V_f \leq 38\%$ ).

**FIG.1**

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**BACKGROUND OF THE INVENTION****FIELD OF THE INVENTION**

5 The present invention relates to a high strength aluminum alloy, and particularly, to an aluminum alloy including an aluminum matrix and a particular intermetallic compound.

**DESCRIPTION OF THE PRIOR ART**

10 There are conventionally known quenched and solidified aluminum alloys as aluminum alloys of this type (for example, see Japanese Patent Application Laid-open No. 275732/89 and U.S. Patent No. 5,053,085).

However, the quenched and solidified aluminum alloys have a problem that they have a high strength, but have a low toughness, on the other hand.

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**SUMMARY OF THE INVENTION**

Accordingly, it is an object of the present invention to provide an aluminum alloy of the type described above, which has a high strength with an increased toughness achieved.

20 To achieve the above object, according to the present invention, there is provided an high strength aluminum alloy, comprising an aluminum matrix, and an intermetallic compound  $Al_3REFe_4$  (wherein RE is at least one element selected from the group consisting of Y, La, Ce, Pr, Nd, Sm, Gd and Dy) having a grain size equal to or less than  $1.0\ \mu m$  and dispersed in said aluminum matrix, the volume fraction  $V_f$  of  $Al_3REFe_4$  being in a range of  $13\% \leq V_f \leq 38\%$ .

25 If  $Al_3REFe_4$  is dispersed in the aluminum matrix as described above, the hardness of the aluminum alloy can be increased to provide an increased strength and to achieve an increase in toughness.

However, if the volume fraction  $V_f$  of  $Al_3REFe_4$  is less than 13% ( $V_f < 13\%$ ), the degree of increase in hardness of the aluminum alloy is low, resulting in a failure to provide an increase in strength. On the other hand, if  $V_f > 38\%$ , the breaking strain  $\epsilon_f$  of the aluminum alloy is extremely low, resulting in a failure to

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achieve an increase in toughness.

The above and other objects, features and advantages of the invention will become apparent from the following detailed description of preferred embodiments, taken in conjunction with the accompanying drawings.

**BRIEF DESCRIPTION OF THE DRAWINGS**

- 35 Fig. 1 is an X-ray diffraction pattern for an aluminum alloy;  
 Fig. 2 is a thermocurve diagram of a differential thermal analysis for the aluminum alloy;  
 Fig. 3 is a graph illustrating the relationship between the time and the temperature in a thermal  
 40 treatment;  
 Fig. 4 is a graph illustrating the relationship between the time  $A$  and the Vickers hardness  $H_v$ ;  
 Fig. 5 is a graph illustrating the relationship between the volume fraction  $V_{f1}$  of  $Al_3Dy$  and the Vickers hardness  $H_v$ ;  
 Fig. 6 is a graph illustrating the relationship between the volume fraction  $V_f$  of  $Al_3DyFe_4$  and the Vickers  
 45 hardness  $H_v$ ;  
 Fig. 7 is a photomicrograph showing the metallographic structure of an aluminum alloy;  
 Fig. 8 is a diagram traced from Fig. 7;  
 Fig. 9 is a graph illustrating the relationship between the volume fraction  $V_{f1}$  of  $Al_3Dy$  and the Vickers hardness  $H_v$ ;  
 50 Fig. 10 is a graph illustrating the relationship between the volume fraction  $V_f$  of  $Al_3DyFe_4$  and the Vickers hardness  $H_v$ ;  
 Fig. 11 is a graph illustrating the relationship between the volume fraction  $V_f$  of  $Al_3REFe_4$  and the Vickers hardness  $H_v$ ;  
 Fig. 12 is a graph illustrating the relationship between the volume fraction  $V_f$  of  $Al_3REFe_4$  and the  
 55 breaking strain  $\epsilon_f$ .  
 Fig. 13 is a graph illustrating the relationship between the grain size of  $Al_3DyFe_4$  and the Vickers hardness  $H_v$ ;  
 Fig. 14 is a graph illustrating the relationship between the Mn content and the Vickers hardness  $H_v$ .

**DESCRIPTION OF THE PREFERRED EMBODIMENTS**

An aluminum alloy according to the present invention is represented by the chemical formula:  $Al_a Fe_b RE_c$ , wherein a, b and c are atomic percentages defined within the following ranges:  $85 \leq a \leq 95$ ,  $3 \leq b \leq 9$ , and  $1 \leq c \leq 6$ . RE is at least one element selected from the group consisting of Y, La, Ce, Pr, Nd, Sm, Gd and Dy.

In producing the aluminum alloy, a process is employed which comprises preparing an aluminum alloy blank having a metallographic structure including an amorphous single-phase texture, or a mixed-phase texture consisting of an amorphous phase and a crystalline phase, and then subjecting the aluminum alloy blank to a thermal treatment. An intermetallic compound  $Al_3 REFe_4$  is precipitated during this thermal treatment step, and dispersed in an aluminum matrix.

A further aluminum alloy according to the present invention is represented by the formula  $Al_a Fe_b RE_c Mn_d$ , wherein a, b, c and d are atomic percentages defined within the following ranges:  $85 \leq a \leq 95$ ,  $2 \leq b \leq 8$ ,  $1 \leq c \leq 6$ , and  $0.5 \leq d \leq 6$ . In this case, an intermetallic compound contributing to an increase in strength of the aluminum alloy is  $Al_3 RE(FeMn)_4$  and has a structure in which a portion of Fe is replaced by Mn.

**Example 1**

A molten metal having a composition represented by  $Al_{92}Fe_5Dy_3$  (each of these numerical values represent atomic percentages) in which Dy is selected as RE was prepared in a high frequency melting furnace. Then, the molten metal was used to produce a ribbon-like aluminum alloy blank having a width of 2 mm and a thickness of about 20  $\mu m$  by application of a single roll process. Conditions for the single roll process were as follows: the diameter of a roll of copper was 250 mm; the number of revolutions of the roll was 4,000 rpm; the diameter of an injection bore in a quartz nozzle was 0.5 mm; a gap between the quartz nozzle and the roll was 0.3 mm; the pressure under which the molten metal was injected was 0.4 kgf/cm<sup>2</sup>; and the atmosphere was an argon atmosphere under -40 cmHg.

Fig. 1 is an X-ray diffraction pattern for the aluminum alloy blank, and Fig. 2 is a thermocurve diagram of a differential thermal analysis (DSC) for the aluminum alloy blank. It can be seen from Figs. 1 and 2 that the aluminum alloy has a mixed-phase texture consisting of an amorphous phase and a crystalline phase. The temperature Tx for crystallization of the amorphous phase is 624.2 K.

The ribbon-like aluminum alloy blank was cut to fabricate a plurality of thin aluminum alloy blank pieces having a length of about 50 mm. These pieces were subjected to a thermal treatment to provide various aluminum alloys.

Fig. 3 shows a thermal treatment process. This process comprises a primary temperature-raising stage P<sub>1</sub> in which the temperature is raised from about 298 K to 590 K at a rate of about 15 K/minute; a primary constant-temperature maintaining stage P<sub>2</sub> in which the temperature is maintained at 590 K for A minutes; a secondary temperature-raising stage P<sub>3</sub> in which the temperature is raised from 590 K to 723 K at a rate of about 9 K/minute; a secondary constant-temperature maintaining stage P<sub>4</sub> in which the temperature is maintained at 723 K for one hour; and a quenching stage P<sub>5</sub> in which the blank is quenched from 723 K. The time for the primary constant-temperature maintaining stage, i.e., A, was varied for every aluminum alloy blank.

Table 1 shows the time A for the primary constant-temperature maintaining stage P<sub>2</sub> for various aluminum alloys (1) to (6); the metallographic structure at the completion of the primary constant-temperature maintaining stage P<sub>2</sub> (i.e., the intermediate metallographic structure); and the metallographic structure at the completion of the quenching stage P<sub>5</sub> (i.e., the final metallographic structure). X-ray diffraction was primarily used to observe these metallographic structures. However, a transmission electron microscope and fluorescent X-ray analysis were also used. In table 1, "amo" means an amorphous phase; "fcc" means an aluminum crystal; and "IMC" means an intermetallic compound, and a "O" mark indicates that an unidentified AlFe and the like are present.

Table 1

Al alloy	Primary constant-temperature maintaining stage P <sub>2</sub>		Final metallographic structure (Dispersed in a Al matrix)		
	A (minute)	I.M.S.	UNID.AIFe	Al <sub>3</sub> DY	Al <sub>8</sub> DyFe <sub>4</sub>
(1)	10	amo + fcc	○	○	-
(2)	30	amo + fcc	○	○	-
(3)	60	fcc + IMC	○	○	○
(4)	90	fcc + IMC	○	○	○
(5)	120	fcc + IMC	-	○	○
(6)	150	fcc + INC	-	○	○
I.M.S. = intermediate metallographic structure					
UNID = unidentified					

The following can be seen from Table 1: In the aluminum alloys (1) and (2), produced with the time A set at 10 and 30 minutes and having the intermediate metallographic structure of "amo + fcc", no Al<sub>8</sub>DyFe<sub>4</sub> was present in the final metallographic structure. In the aluminum alloys (3) and (4), produced with the time A set at 60 and 90 minutes and having the intermediate metallographic structure of "fcc + IMC", unidentified AlFe, Al<sub>3</sub>Dy and Al<sub>8</sub>DyFe<sub>4</sub> were observed in the final metallographic structure. Further, in the aluminum alloys (5) and (6), produced with the time A set at 120 and 150 minutes and having the intermediate metallographic structure of "fcc + IMC", Al<sub>3</sub>Dy and Al<sub>8</sub>DyFe<sub>4</sub> were observed in the final metallographic structure. These facts indicate that the configuration of the final metallographic structure is varied depending upon the intermediate metallographic structure.

Fig. 4 shows the relationship between the time A and the Vickers hardness Hv, wherein characters (1) to (6) correspond to the aluminum alloys (1) to (6).

It can be seen from Fig. 4 that the longer the time A is, the higher the hardness is. This phenomenon is attributable to the difference in IMC and the amount of IMC present, because there is less difference in size of the metallographic structure between the aluminum alloys (1) to (6).

Thereupon, the volume fractions Vf and Vf<sub>1</sub> of Al<sub>8</sub>DyFe<sub>4</sub> and Al<sub>3</sub>Dy in each of the aluminum alloys (1) to (6) were determined from a scanning electron microscope photomicrograph (a transmission electron microscope photomicrograph and a fluorescent X-ray analysis photograph were also partially used) to provide the results given in Table 2.

Table 2

Al alloy	Vf <sub>1</sub> of Al <sub>3</sub> Dy (%)	Vf of Al <sub>8</sub> DyFe <sub>4</sub> (%)
(1)	13.5	--
(2)	13.5	--
(3)	11	9
(4)	9	16
(5)	8	19
(6)	7	20

It can be seen from Table 2, that if the volume fraction Vf<sub>1</sub> of Al<sub>3</sub>Dy is decreased, the volume fraction Vf of the Al<sub>8</sub>DyFe<sub>4</sub> is increased.

Fig. 5 shows the relationship between the volume fraction Vf<sub>1</sub> of Al<sub>3</sub>Dy and the Vickers hardness of each of the aluminum alloys (1), (3) to (6), and Fig. 6 shows the relationship between the volume fraction Vf of Al<sub>8</sub>DyFe<sub>4</sub> and the Vickers hardness of each of the aluminum alloys (1), (3) to (6). Characters (1), (3) to (6) in Figs. 5 and 6 correspond to the aluminum alloys (1), (3) to (6), respectively. It can be seen from Figs. 5 and 6 that the Vickers hardness of the aluminum alloy is increased by a decrease in volume fraction Vf<sub>1</sub> of the Al<sub>3</sub>Dy and by an increase in volume fraction Vf of Al<sub>8</sub>DyFe<sub>4</sub>. In this case, the grain size of Al<sub>8</sub>DyFe<sub>4</sub> was in a range of 100 to 300 nm.

Fig. 7 shows a scanning electron microscope photomicrograph (72,000X magnification) of the metallographic structure of the aluminum alloy (6). Fig. 8 is a diagram traced from Fig. 8. In Figs. 7 and 8, the

number of  $\text{Al}_3\text{Dy}$  is 19, and the remaining is  $\text{Al}_8\text{DyFe}_4$ . It can be seen from both of Figs. 7 and 8 that these  $\text{Al}_8\text{DyFe}_4$  are dispersed in the aluminum matrix M.

Various aluminum alloys (7) to (15) of the same type as the above-described aluminum alloys were produced in the same manner as described above.

Table 3 shows the composition of the various aluminum alloys (7) to (15) and the time A taken for the primary constant-temperature maintaining stage  $P_2$ .

Table 3

Al alloy	Composition (by atomic %)	Time A (minute)
(7)	$\text{Al}_{91}\text{Fe}_6\text{Dy}_3$	60
(8)	$\text{Al}_{91}\text{Fe}_6\text{Dy}_3$	90
(9)	$\text{Al}_{91}\text{Fe}_6\text{Dy}_3$	120
(10)	$\text{Al}_{90}\text{Fe}_7\text{Dy}_3$	60
(11)	$\text{Al}_{90}\text{Fe}_7\text{Dy}_3$	90
(12)	$\text{Al}_{90}\text{Fe}_7\text{Dy}_3$	120
(13)	$\text{Al}_{89}\text{Fe}_8\text{Dy}_3$	90
(14)	$\text{Al}_{89}\text{Fe}_8\text{Dy}_3$	120
(15)	$\text{Al}_{90}\text{Fe}_8\text{Dy}_2$	90

Fig. 9 shows their relationship between the volume fraction  $V_f$  of  $\text{Al}_3\text{Dy}$  and the Vickers hardness  $H_v$  of each of the aluminum alloys (7) to (15), and Fig. 10 shows the relationship between the volume fraction  $V_f$  of  $\text{Al}_8\text{DyFe}_4$  and the Vickers hardness of each of the aluminum alloys (7) to (15). Characters (7) to (15) in Figs. 9 and 10 correspond to the aluminum alloys (7) to (15), respectively.

It was confirmed from Figs. 9 and 10 that even if the composition of the aluminum alloy was changed, the Vickers hardness  $H_v$  of the aluminum alloy is increased by a decrease in volume fraction  $V_f$  of  $\text{Al}_3\text{Dy}$  and an increase in volume fraction  $V_f$  of  $\text{Al}_8\text{DyFe}_4$ .

Further, various aluminum alloys (16) to (25) with the type of RE varied were produced by the same process as described above.

Table 4 shows the composition of each of the various aluminum alloys (16) to (25), the time A taken for the primary constant-temperature maintaining stage  $P_2$  and intermetallic compounds represented by  $\text{Al}_8\text{REFe}_4$ .

Table 4

Al alloy	Composition (by atomic %)	Time A (minute)	$\text{Al}_8\text{REFe}_4$
(16)	$\text{Al}_{92}\text{Fe}_5\text{Y}_3$	60	$\text{Al}_8\text{YFe}_4$
(17)	$\text{Al}_{92}\text{Fe}_5\text{Y}_3$	120	$\text{Al}_8\text{YFe}_4$
(18)	$\text{Al}_{90}\text{Fe}_6\text{La}_2\text{Ce}_2$	90	$\text{Al}_8(\text{LaCe})\text{Fe}_4$
(19)	$\text{Al}_{90}\text{Fe}_6\text{La}_2\text{Ce}_2$	120	$\text{Al}_8(\text{LaCe})\text{Fe}_4$
(20)	$\text{Al}_{91}\text{Fe}_5\text{Nd}_2\text{Pr}_2$	90	$\text{Al}_8(\text{NdPr})\text{Fe}_4$
(21)	$\text{Al}_{91}\text{Fe}_5\text{Nd}_2\text{Pr}_2$	120	$\text{Al}_8(\text{NdPr})\text{Fe}_4$
(22)	$\text{Al}_{90}\text{Fe}_6\text{Sm}_2\text{Gd}_2$	90	$\text{Al}_8(\text{SmGd})\text{Fe}_4$
(23)	$\text{Al}_{90}\text{Fe}_6\text{Sm}_2\text{Gd}_2$	120	$\text{Al}_8(\text{SmGd})\text{Fe}_4$
(24)	$\text{Al}_{89}\text{Fe}_7\text{La}_2\text{Ce}_1\text{Nd}_1$	90	$\text{Al}_8(\text{LaCeNd})\text{Fe}_4$
(25)	$\text{Al}_{89}\text{Fe}_7\text{La}_2\text{Ce}_1\text{Nd}_1$	120	$\text{Al}_8(\text{LaCeNd})\text{Fe}_4$

Fig. 11 shows the relationship between the volume fraction  $V_f$  of  $\text{Al}_8\text{REFe}_4$  and the Vickers hardness  $H_v$  in each of the various aluminum alloys (16) to (25) and the above-described aluminum alloys (1) to (15). Characters (1) to (25) in Fig. 11 correspond to the aluminum alloys (1) to (25), respectively.

As is apparent from Fig. 11, in the various aluminum alloys, the increase rate of the Vickers hardness  $H_v$  is increased at the volume fraction  $V_f$  of  $\text{Al}_8\text{REFe}_4$  equal to or more than 13%. Therefore, the lower limit

value of the volume fraction  $V_f$  of  $Al_8REFe_4$  required for increasing the strength of the aluminum alloy is set at 13% ( $V_f = 13\%$ ).

Then, the breaking strain  $\epsilon_f$  of each of the aluminum alloys (1) to (25) was measured in order to examine the toughness of each of these aluminum alloys.

Fig. 12 shows the relationship between the volume fraction  $V_f$  of  $Al_8REFe_4$  and the breaking strain  $\epsilon_f$  in each of the various aluminum alloys (1) to (25). Characters (1) to (25) in Fig. 12 correspond to the aluminum alloys (1) to (25), respectively.

As is apparent from Fig. 12, if the volume fraction  $V_f$  of  $Al_8REFe_4$  becomes more than 38%, the breaking strain  $\epsilon_f$  is equal to or less than 0.01, resulting in a failure to achieve an increase in toughness of the aluminum alloy. Therefore, the upper limit value of the volume fraction  $V_f$  of  $Al_8REFe_4$  is set at 38% ( $V_f = 38\%$ ).

The grain size of  $Al_8REFe_4$  will be described below.

A plurality of thin aluminum alloy blank pieces having compositions similar to those described above, i.e.,  $Al_{92}Fe_5Dy_3$  (each numerical value represents an atomic %) were subjected to the primary constant-temperature maintaining stage  $P_2$  with the time  $A$  set at 120 minutes ( $A = 120$  minutes) and with the temperature likewise set at 590 K, thereby producing intermediate products which were then subjected to the secondary constant-temperature maintaining stage  $P_1$  with the temperature likewise set at 723 to 823 K and with the time set at 1 to 30 hours, thereby producing various aluminum alloys (26) to (30) having different grain sizes of  $Al_8DyFe_4$ . Conditions established for the other stages  $P_1$ ,  $P_3$  and  $P_5$  were the same as those described with reference to Fig. 3.

Table 5 shows the volume fractions  $V_f$  of  $Al_8DyFe_4$  in the various aluminum alloys (26) to (30).

Table 5

Al alloy	$V_f$ of $Al_8DyFe_4$ (%)
(26)	20
(27)	19
(28)	19
(29)	20
(30)	20

Fig. 13 shows the relationship between the grain size of  $Al_8DyFe_4$  and the Vickers hardness  $H_v$  of each of the aluminum alloys (26) to (30). Characters (26) to (30) in Fig. 13 correspond to the aluminum alloys (26) to (30), respectively. As is apparent from Fig. 13, the Vickers hardness  $H_v$  of the aluminum alloy can be increased by setting the grain size of  $Al_8DyFe_4$  less than  $1.0 \mu m$ , thereby increasing the strength of the aluminum alloy.

## Example 2

Various molten metals having compositions obtained by replacement of all or a portion of Fe in  $Al_{92}Fe_5Dy_3$  (each numerical value represents an atomic %) by Mn were prepared in a high frequency melting furnace and then subjected to a single-roll process under the same conditions as in Example 1 to produce various aluminum alloy blanks.

Table 6 shows the compositions of the various aluminum alloy blanks (31) to (35).

Table 6

Al alloy blank	Composition (by atomic %)
(31)	$Al_{92}Fe_4Dy_3Mn_1$
(32)	$Al_{92}Fe_3Dy_3Mn_2$
(33)	$Al_{92}Fe_2Dy_3Mn_3$
(34)	$Al_{92}Fe_1Dy_3Mn_4$
(35)	$Al_{92}Dy_3Mn_5$

An X-ray diffraction and a differential thermal analysis (DSC) showed that each of the aluminum alloy blanks (31) to (34) had a mixed-phase texture consisting of an amorphous phase and a crystalline phase, but the aluminum alloy blank (35) had a crystalline single-phase texture due to its composition. In addition, each of the aluminum alloy blanks (31) to (34) was formed into a ribbon-like configuration and exhibited a toughness, whereas the aluminum alloy blank (35) was also formed into a ribbon-like configuration, but was very brittle.

Then, except for the aluminum alloy blank (35), the other aluminum alloy blanks (31) to (34) were subjected to a thermal treatment to produce aluminum alloys (31) to (34) (for convenience, the same characters are used for the aluminum alloys corresponding to the aluminum alloy blanks). Conditions for the thermal treatment are similar to those described with reference to Fig. 3, except for the time A set at 120 minutes for the primary constant-temperature maintaining stage P<sub>2</sub>.

Intermetallic compounds in the aluminum alloys (31) to (34) are Al<sub>3</sub>Dy and Al<sub>8</sub>Dy(FeMn)<sub>4</sub>, and the volume fractions Vf<sub>3</sub> of Al<sub>8</sub>Dy(FeMn)<sub>4</sub> in the aluminum alloys (31) to (34) are as shown in Table 7.

Table 7

Al alloy blank	Vf <sub>3</sub> of Al <sub>8</sub> Dy(FeMn) <sub>4</sub> (%)
(31)	19
(32)	19
(33)	20
(34)	19

It can be seen from Table 7 that the volume fractions Vf<sub>3</sub> of Al<sub>8</sub>Dy(FeMn)<sub>4</sub> in the aluminum alloys (31) to (34) are in a range of  $13\% \leq Vf_3 \leq 38\%$ .

Fig. 14 shows the relationship between the amount of added Mn and the Vickers hardness Hv in each of the aluminum alloys (31) to (34). Characters (31) to (34) in Fig. 14 correspond to the aluminum alloys (31) to (34), respectively. A character (5) corresponds to the above-described aluminum alloy (5).

As is apparent from Fig. 14, each of the aluminum alloys (31) to (34) having compositions obtained by replacement of all or a portion of Fe in Al<sub>8</sub>DyFe<sub>4</sub> has an increased Vickers hardness Hv and thus an increased strength, as compared with those of the aluminum alloy (5) free from Mn. This is believed to be attributable to an increased resistance to deformation in a crystal grain boundary as a result of addition of Mn.

#### Claims

1. A high strength aluminum alloy, comprising an aluminum matrix, and an intermetallic compound Al<sub>8</sub>REFe<sub>4</sub>, wherein:
  - a. RE is at least one element selected from the group consisting of Y, La, Ce, Pr, Nd, Sm, Gd, and Dy;
  - b. said intermetallic compound has a grain size less than or equal to 1.0 μm;
  - c. said intermetallic compound is dispersed in said aluminum matrix; and
  - d. the volume fraction Vf of said intermetallic compound is in the following range:  $13\% \leq Vf \leq 38\%$ .
2. A high strength aluminum alloy according to claim 1, wherein the aluminum alloy is represented by the chemical formula Al<sub>a</sub>Fe<sub>b</sub>RE<sub>c</sub>, and wherein a, b, and c are atomic percentages within the following ranges:
 

$85 \leq a \leq 95$ ;  
 $3 \leq b \leq 9$ ; and  
 $1 \leq c \leq 6$ .
3. A high strength aluminum alloy, comprising an aluminum matrix, and an intermetallic compound Al<sub>8</sub>RE-(FeMn)<sub>4</sub>, wherein:
  - a. RE is at least one element selected from the group consisting of Y, La, Ce, Pr, Nd, Sm, Gd, and Dy;
  - b. said intermetallic compound has a grain size less than or equal to 1.0 μm;

c. said intermetallic compound is dispersed in said aluminum matrix; and

d. the volume fraction  $V_f$  of said intermetallic compound is in the following range:  $13\% \leq V_f \leq 38\%$ .

4. A high strength aluminum alloy according to claim 3, wherein the aluminum alloy is represented by the chemical formula  $Al_aFe_bRE_cMn_d$ , and wherein a, b, c, and d are atomic percentages within the following ranges:

$85 \leq a \leq 95$ ;

$2 \leq b \leq 8$ ;

$1 \leq c \leq 6$ ; and

$0.5 \leq d \leq 6$ .

5. A method for producing a high strength aluminum alloy having an intermetallic compound  $Al_3REFe_4$  dispersed in an aluminum matrix, wherein RE is at least one element selected from the group consisting of Y, La, Ce, Pr, Nd, Sm, Gd, and Dy; the grain size of the intermetallic compound is equal to or less than  $1.0 \mu m$ ; and the volume fraction  $V_f$  of the intermetallic is in a range of  $13\% \leq V_f \leq 38\%$ , said method comprising:

a. producing an aluminum alloy blank having a metallographic structure selected from the group consisting of an amorphous single phase texture and a mixed-phase texture consisting of an amorphous phase and a crystalline phase;

b. subjecting said aluminum alloy blank to a thermal treatment to precipitate the intermetallic compound.

6. A method for producing a high strength aluminum alloy according to claim 5, wherein said thermal treatment comprises:

a. raising the temperature of said aluminum alloy blank from about ambient temperature to about 590 K at a rate of approximately 15 K/minute;

b. maintaining the temperature of said aluminum alloy blank at a temperature of about 590 K for a period sufficient to precipitate the intermetallic compound, and thereby producing an intermediate microstructure;

c. increasing the temperature of said aluminum alloy from about 590 K to a temperature in the range of about 723 K to 823 K at a rate of about 9 K/minute;

d. maintaining the temperature of said aluminum alloy in the temperature range of about 723 K to about 823 K for a period of approximately 1 hour to at most the period of time required to increase the grain size of the intermetallic phase to  $1.0 \mu m$ ;

e. quenching the aluminum alloy and thereby fixing a final metallographic structure in the aluminum alloy.

7. A method for producing a high strength aluminum alloy according to claim 5 or 6, wherein the aluminum alloy is represented by the chemical formula  $Al_aFe_bRE_c$ , and wherein a, b, and c are atomic percentages within the following ranges:

$85 \leq a \leq 95$ ;

$3 \leq b \leq 9$ ; and

$1 \leq c \leq 6$ .

8. A method for producing a high strength aluminum alloy having an intermetallic compound  $Al_3RE(FeMn)_4$  dispersed in an aluminum matrix, wherein RE is at least one element selected from the group consisting of Y, La, Ce, Pr, Nd, Sm, Gd, and Dy; the grain size of the intermetallic compound is equal to or less than  $1.0 \mu m$ ; and the volume fraction  $V_f$  of the intermetallic is in a range of  $13\% \leq V_f \leq 38\%$ , said method comprising:

a. producing an aluminum alloy blank having a metallographic structure selected from the group consisting of an amorphous single phase texture and a mixed-phase texture consisting of an amorphous phase and a crystalline phase;

b. subjecting said aluminum alloy blank to a thermal treatment to precipitate the intermetallic compound.



9. A method for producing a high strength aluminum alloy according to claim 8, wherein said thermal treatment comprises:

a. raising the temperature of said aluminum alloy blank from about ambient temperature to about 590 K at a rate of approximately 15 K/minute;

b. maintaining the temperature of said aluminum alloy blank at a temperature of about 590 K for a period sufficient to precipitate the intermetallic compound, and thereby producing an intermediate microstructure;

c. Increasing the temperature of said aluminum alloy from about 590 K to a temperature in the range of about 723 K to 823 K at a rate of about 9 K/minute;

d. maintaining the temperature of said aluminum alloy in the temperature range of about 723 K to about 823 K for a period of approximately 1 hour to at most the period of time required to increase the grain size of the intermetallic phase to 1.0  $\mu\text{m}$ ;

e. quenching the aluminum alloy and thereby fixing a final metallographic structure in the aluminum alloy.

10. A high strength aluminum alloy according to claim 8 or 9, wherein the aluminum alloy is represented by the chemical formula  $\text{Al}_a\text{Fe}_b\text{RE}_c\text{Mn}_d$ , and wherein a, b, c, and d are atomic percentages within the following ranges:

$85 \leq a \leq 95$ ;

$2 \leq b \leq 8$ ;

$1 \leq c \leq 6$ ; and

$0.5 \leq d \leq 6$ .

FIG.1

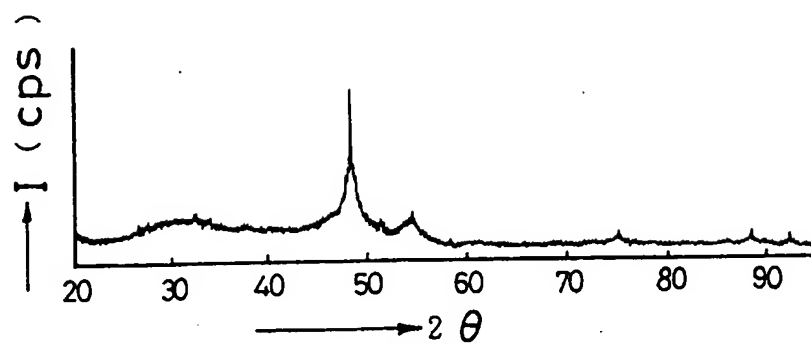


FIG.2

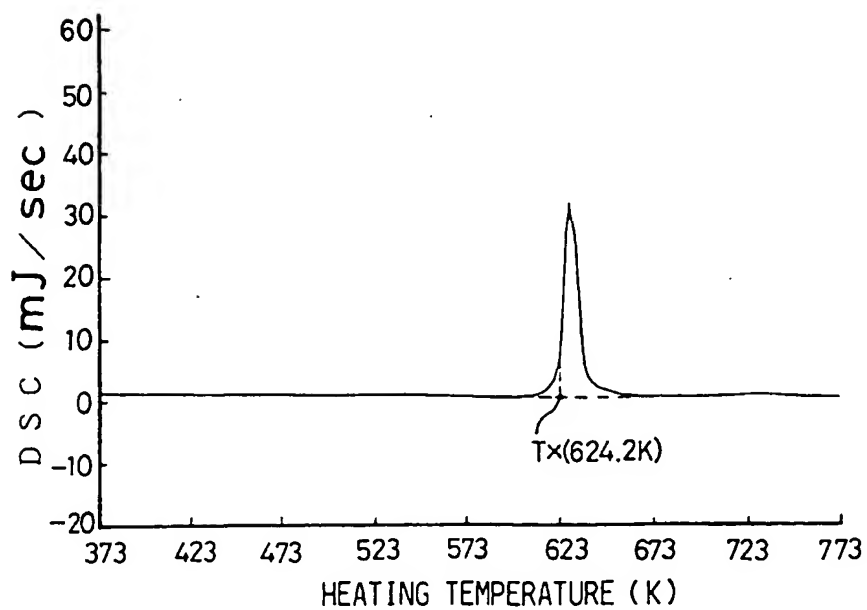


FIG.3

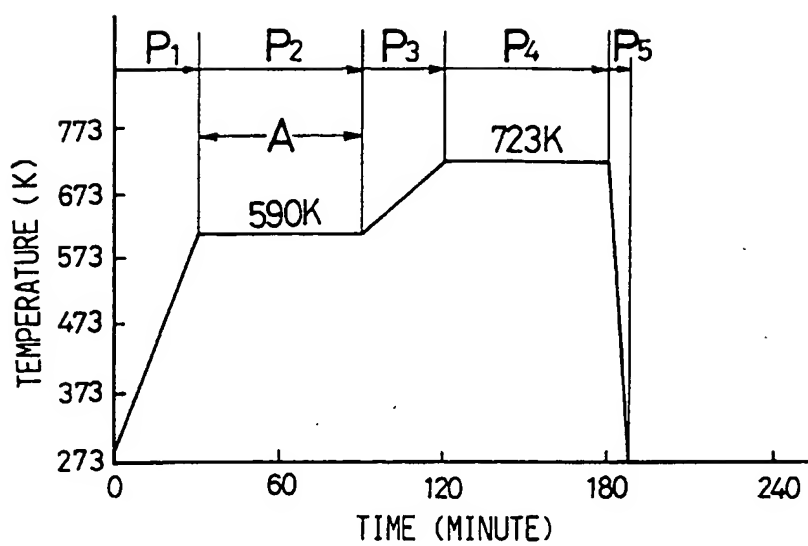


FIG.4

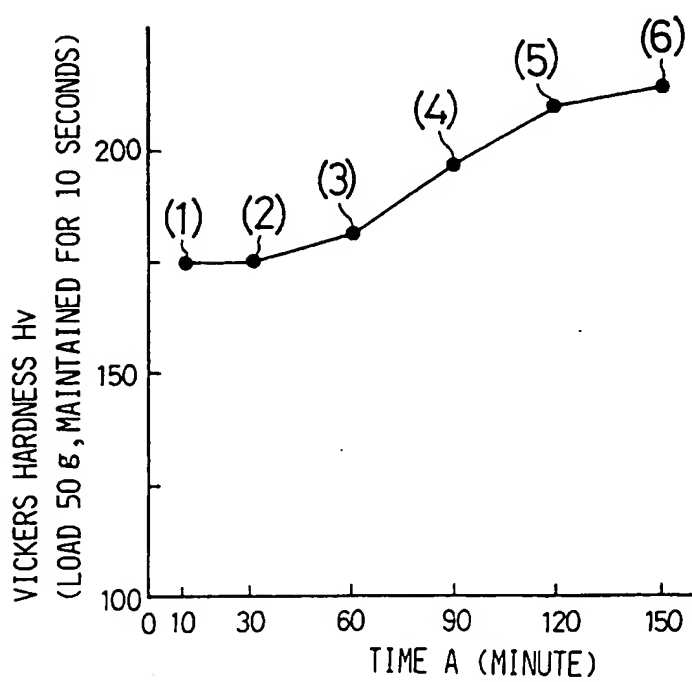


FIG.5

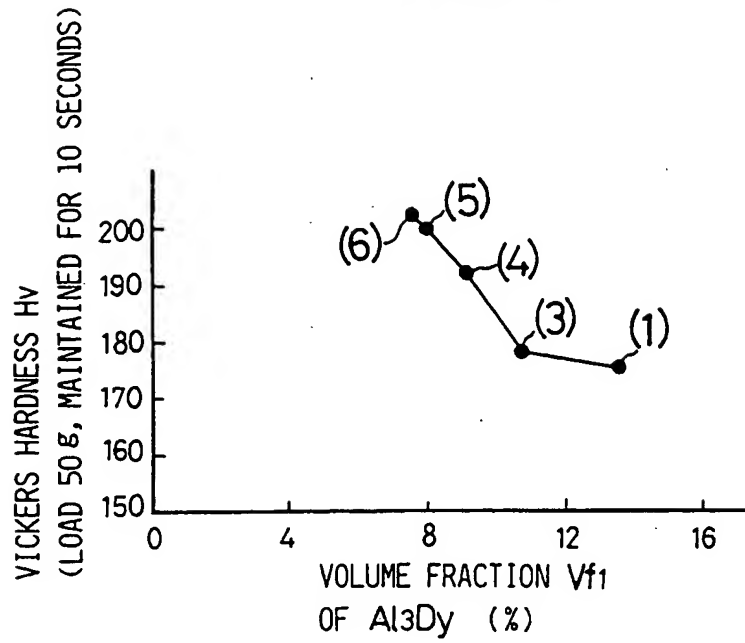


FIG.6

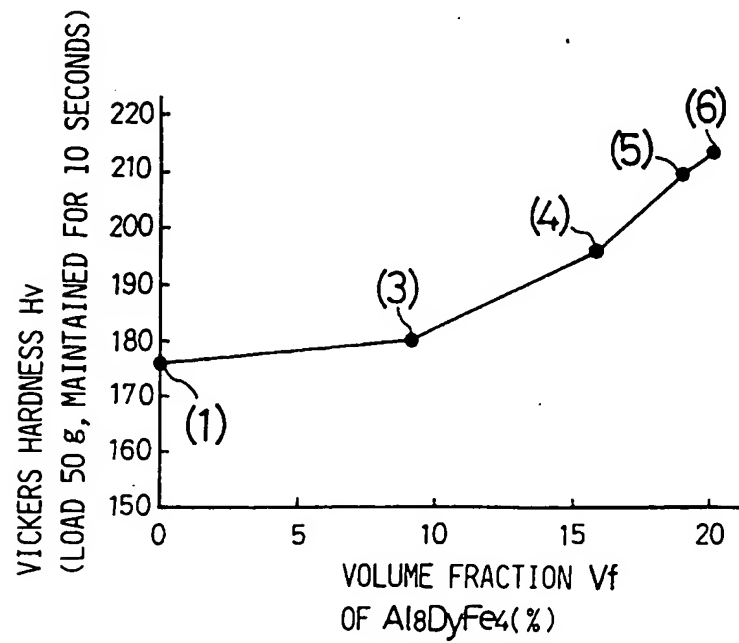


FIG.7

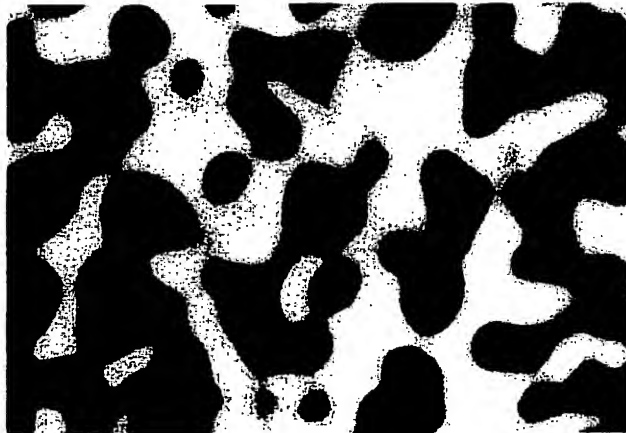


FIG. 8

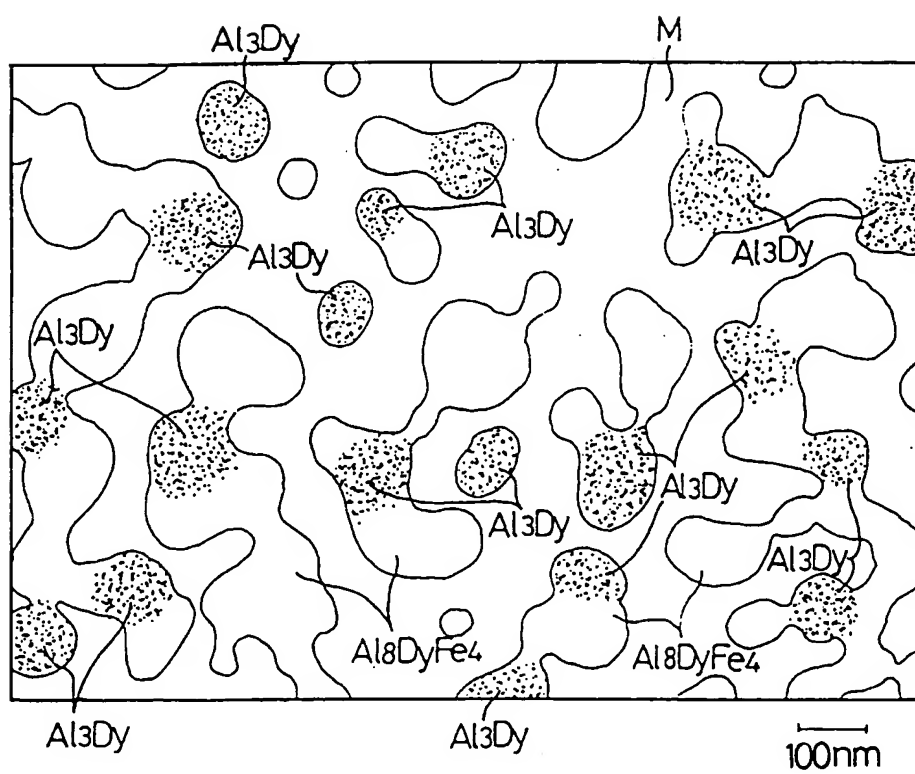


FIG.9

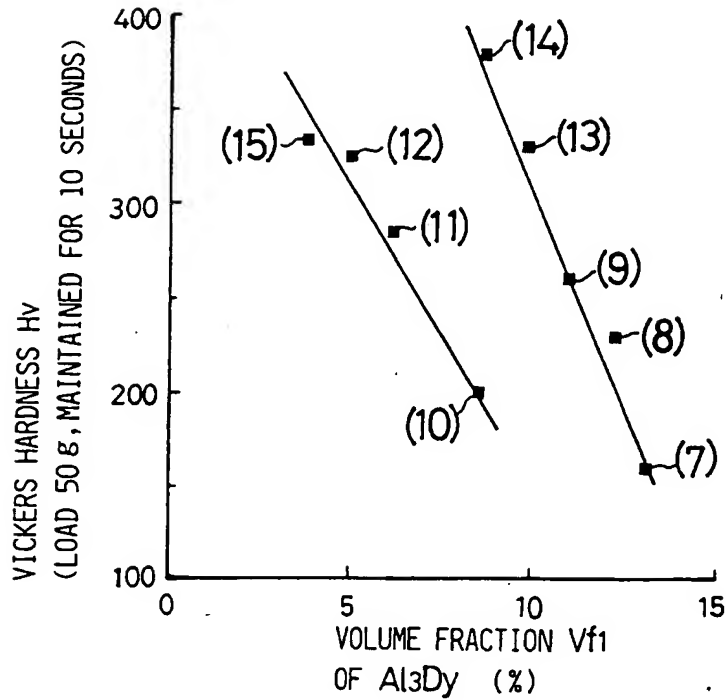


FIG.10

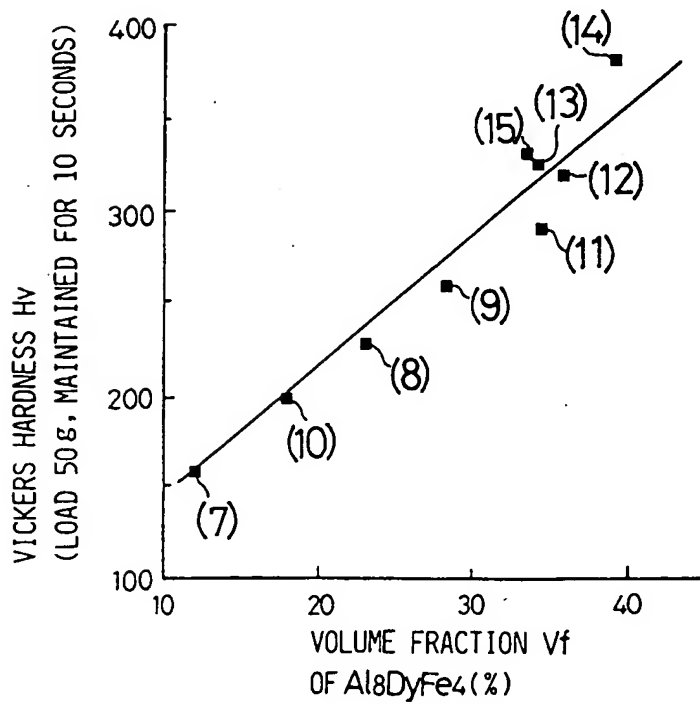


FIG.11

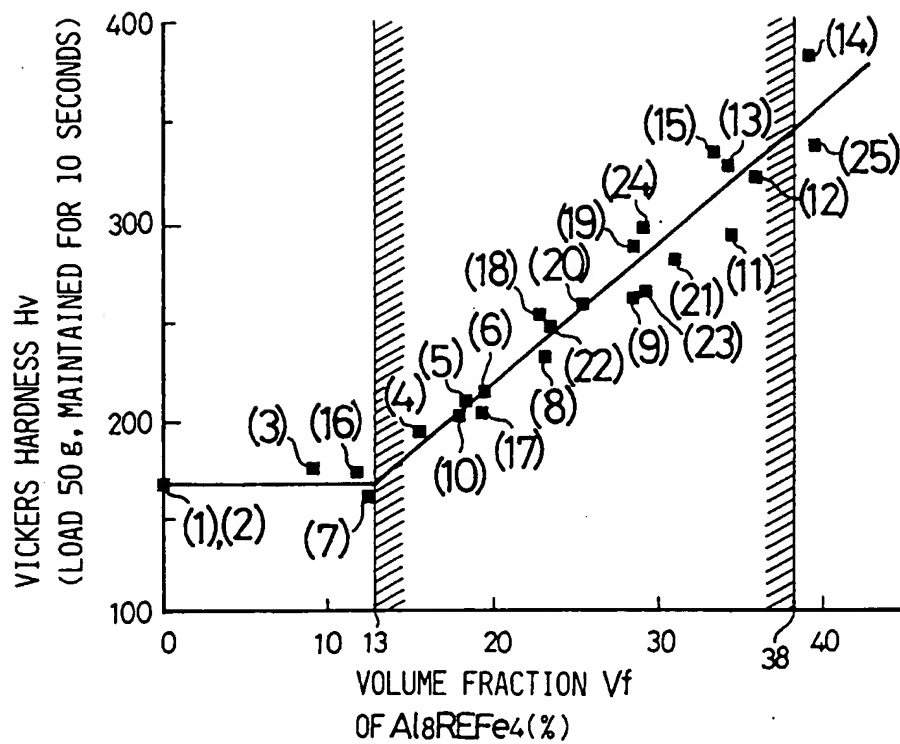




FIG.12

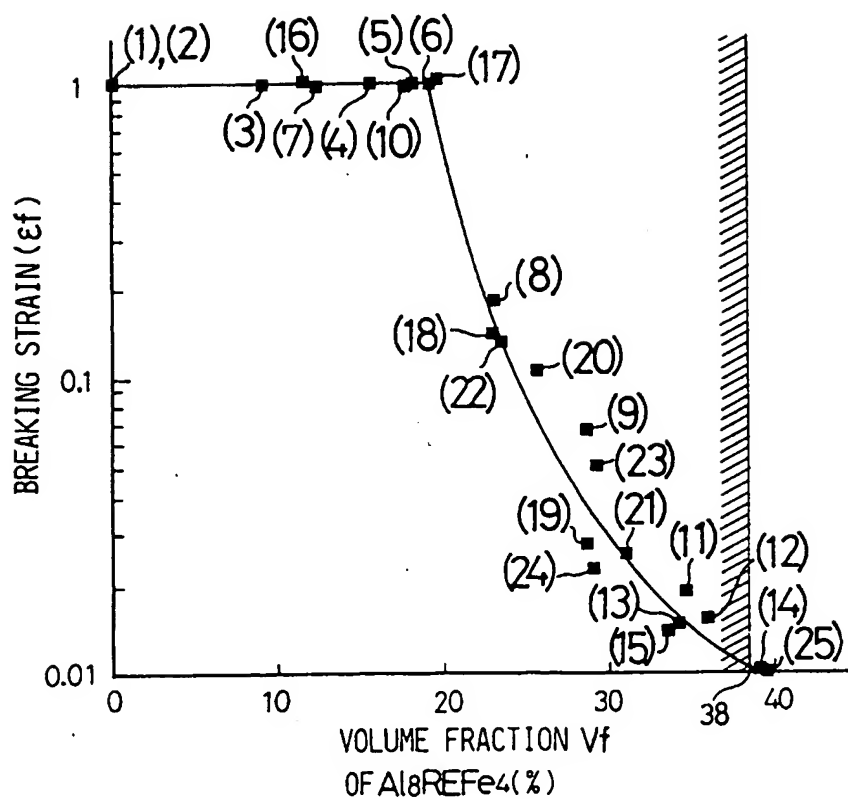


FIG.13

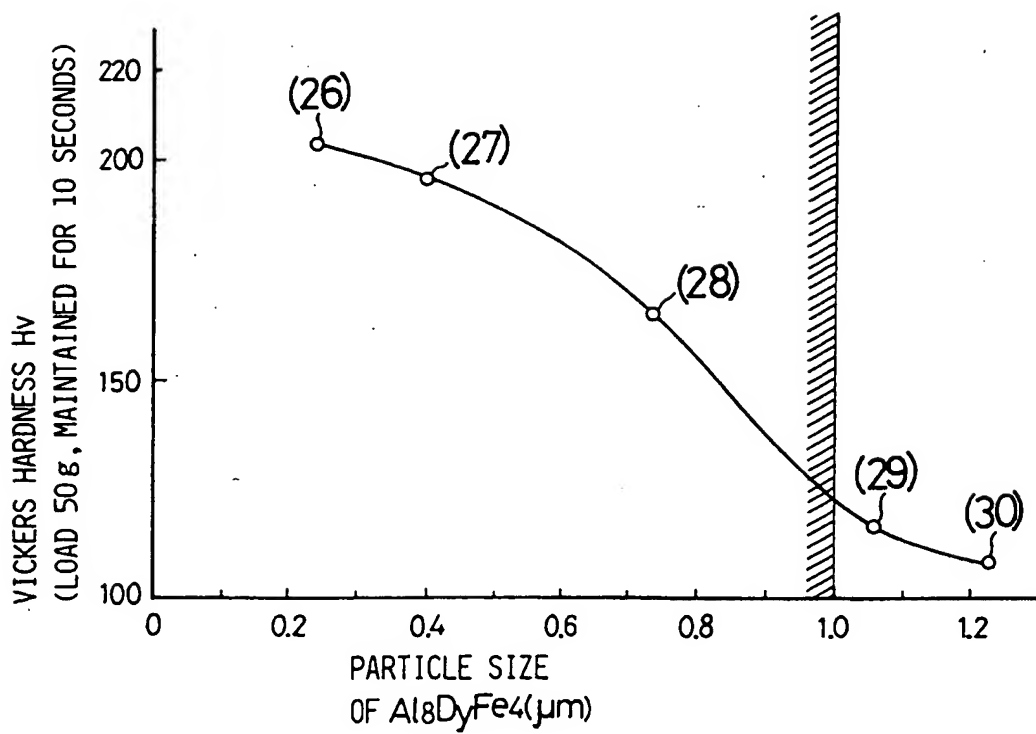
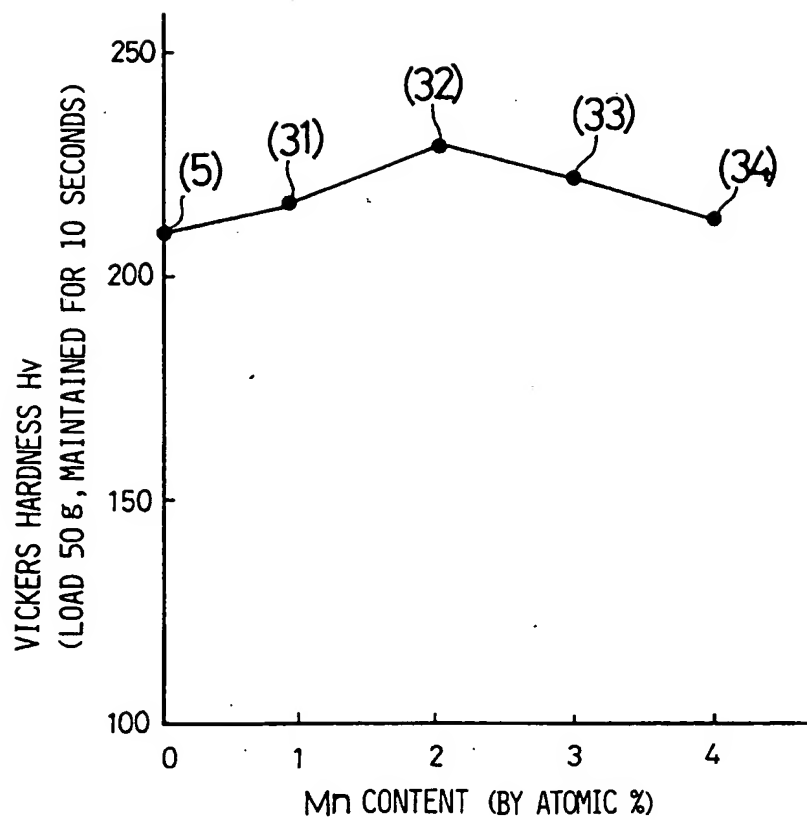


FIG.14





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## EUROPEAN SEARCH REPORT

Application Number

EP 93 10 8092

DOCUMENTS CONSIDERED TO BE RELEVANT			
Category	Citation of document with indication, where appropriate, of relevant passages	Relevant to claim	CLASSIFICATION OF THE APPLICATION (Int. Cl.5)
X	EP-A-0 136 508 (ALLIED CORPORATION) 10 April 1985 *table 3, page 11, lines 25-30, page 12, lines 5-11* ---	1,2,5,7	C22C21/00 C22C45/08 C22F1/00
A	EP-A-0 339 676 (YOSHIDA KOGYO K.K.) 2 January 1989 ---	1-10	
A	EP-A-0 333 216 (YOSHIDA KOGYO K.K.) 20 September 1989 ---		
A	EP-A-0 475 101 (YOSHIDA KOGYO K.K.) 18 March 1992 ---		
A	EP-A-0 460 887 (YOSHIDA KOGYO K.K.) 11 December 1991 -----		
			TECHNICAL FIELDS SEARCHED (Int. Cl.5)
			C22C C22F
The present search report has been drawn up for all claims			
Place of search MUNICH		Date of completion of the search 11 AUGUST 1993	Examiner G. BADCOCK
CATEGORY OF CITED DOCUMENTS X : particularly relevant if taken alone Y : particularly relevant if combined with another document of the same category A : technological background O : non-written disclosure P : intermediate document		T : theory or principle underlying the invention E : earlier patent document, but published on, or after the filing date D : document cited in the application L : document cited for other reasons ----- A : number of the same patent family, corresponding document	

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